

AD-A123 893

IDEAL ELASTIC ANELASTIC AND VISCOELASTIC FLOW IN A
METALLIC GLASS(U) HARVARD UNIV CAMBRIDGE MA DIV OF
APPLIED SCIENCES R I TRUB ET AL JUL 82 TR-15

1/1

UNCLASSIFIED

N00014-77-C-0002

F/G 20/12

NL



END
FILED
OHC

Office of Naval Research

Contract N00014-77-C-0002 NR-039-136



Accession For		
NTIS GRA&I	<input checked="" type="checkbox"/>	<input type="checkbox"/>
DTIC TAB	<input type="checkbox"/>	<input type="checkbox"/>
Unannounced	<input type="checkbox"/>	<input type="checkbox"/>
Justification		
By		
Distribution/		
Availability Cases		
Dist	Avail end/or Special	
A		

IDEAL ELASTIC, ANELASTIC, AND
VISCOELASTIC FLOW IN A METALLIC GLASS

By

A.I. Taub and F. Spaepen

TR 15

Technical Report No. 15

This document has been approved for public release and sale; its distribution is unlimited. Reproduction in whole or in part is permitted by the U. S. Government.

July 1982

The research reported in this document was made possible through support extended the Division of Applied Sciences, Harvard University, by the Office of Naval Research, under Contract N00014-77-C-0002.

Division of Applied Sciences

Harvard University • Cambridge, Massachusetts

UNCLASSIFIED

SECURITY CLASSIFICATION OF THIS PAGE (When Data Entered)

REPORT DOCUMENTATION PAGE		READ INSTRUCTIONS BEFORE COMPLETING THIS FORM
1. REPORT NUMBER	2. GOVT ACCESSION NO.	3. RECIPIENT'S CATALOG NUMBER
Technical Report No. 15	AD-A123 893.	
4. TITLE (and Subtitle)	5. TYPE OF REPORT & PERIOD COVERED	
IDEAL ELASTIC, ANELASTIC, AND VISCOELASTIC FLOW IN A METALLIC GLASS	Interim Report	
7. AUTHOR(s)	6. PERFORMING ORG. REPORT NUMBER	
A.I. Taub F. Spaepen	N00014-77-C-0002	
9. PERFORMING ORGANIZATION NAME AND ADDRESS	10. PROGRAM ELEMENT, PROJECT, TASK AREA & WORK UNIT NUMBERS	
Division of Applied Sciences Harvard University Cambridge, MA 02138		
11. CONTROLLING OFFICE NAME AND ADDRESS	12. REPORT DATE	
	July 1982	
	13. NUMBER OF PAGES	
	5	
14. MONITORING AGENCY NAME & ADDRESS (if different from Controlling Office)	15. SECURITY CLASS. (of this report)	
	Unclassified	
	16a. DECLASSIFICATION/DOWNGRADING SCHEDULE	
16. DISTRIBUTION STATEMENT (of this Report)	This document has been approved for public release and sale; its distribution is unlimited. Reproduction in whole or in part is permitted by the U.S. Government	
17. DISTRIBUTION STATEMENT (of the abstract entered in Block 20, if different from Report)		
18. SUPPLEMENTARY NOTES		
19. KEY WORDS (Continue on reverse side if necessary and identify by block number)	amorphous alloy, creep, structural relaxation, viscosity, anelasticity, viscoelasticity, PdSi alloy, elasticity	
20. ABSTRACT (Continue on reverse side if necessary and identify by block number)	The elastic, viscoelastic, and anelastic components of the homogeneous strain response of the metallic glass $Pd_{82}Si_{18}$ to an applied stress have been examined. The elastic response is fully reversible, instantaneous, and linear. The measured elastic modulus and temperature dependence are $E = 84 \pm 8$ GPa and $d(\ln E)/dt = (-3.2 \pm 0.6) \times 10^{-3}^{\circ}C$. This viscoelastic flow is nonrecoverable, and, if the configuration remains constant, is characterized by a constant strain rate. This strain rate varies linearly with the stress in the low stress regime ($\gamma < 300$ MPa).	

UNCLASSIFIED

SECURITY CLASSIFICATION OF THIS PAGE(When Data Entered)

becoming nonlinear for higher stresses. For isoconfigurational flow, the strain rate has an Arrhenius-type temperature dependence with an activation energy of -200 \pm 15 kJ/mole, independent of stress and thermal history. The magnitude of the strain rate is strongly dependent on the degree of structural relaxation and therefore on thermal history. During isothermal annealing, the viscoelastic strain rate varies inversely with time.

The anelastic response is a transient that at 500 K contributes to the flow for approximately 50 hours after a stress increase and is fully recovered upon stress reduction. A spectrum of exponential decays is required to model this flow component. The anelastic strain varies linearly with the magnitude of the stress change over the entire stress range tested, [$\gamma_A/\Delta\tau = (8.0 \pm 0.80) \times 10^{-6}$ cm/cm-MPa].

SECURITY CLASSIFICATION OF THIS PAGE(When Data Entered)

IDEAL ELASTIC, ANELASTIC, AND VISCOELASTIC FLOW IN A METALLIC GLASS

A.I. Taub and F. Spaepen*

INTRODUCTION

When a material is subjected to an applied stress, its strain response generally consists of several components, each differing by its dependence on time, by the degree to which it is recovered upon removal of the stress, and by the linearity of the response.⁽¹⁾ Glassy metallic alloy systems have been reported to exhibit at least four of these strain components:

1. Ideal elasticity (recoverable, instantaneous, linear stress-strain)^(2,3)
2. Anelasticity (recoverable, time-dependent, linear stress-strain)⁽⁴⁻⁸⁾
3. Viscoelasticity (permanent, time-dependent, linear stress-strain rate)^(5,6,9,10)
4. Instantaneous plasticity (permanent, instantaneous, nonlinear stress-strain)⁽¹¹⁾

In this study, an amorphous Pd-based alloy was tested in the homogeneous flow regime. This eliminates the strain contribution of the instantaneous plastic flow mechanism.^(12,13) For each of the three remaining strain components, the time dependence, recoverability, and linearity are examined in detail. The effect of structural relaxation on each type of strain is also discussed.

RESULTS

The tensile creep flow of amorphous $Pd_{48}Si_{18}$ wires was measured by the technique described previously.⁽⁹⁾ Both as-quenched and preannealed specimens were tested. The preanneals served to stabilize the structure of the specimens, thus permitting isoconfigurational testing. For preannealed specimens that were to undergo stress reductions during the creep test, load cycling was conducted during the anneal in order to eliminate any structural relaxation that might be associated with load removal during the test.⁽¹⁴⁾

Figure 1a shows a typical strain vs time plot obtained for a preannealed specimen after a stress increase. The preanneal in this case was 325 hours at 500 K. The instantaneous elongation γ_E is characterized by ideal elasticity. The steady-state strain γ_p , established after approximately 50 hours, is viscoelastic. The anelastic strain contribution γ_A is the transient flow that occurs before the steady-state condition is established.

*Harvard University

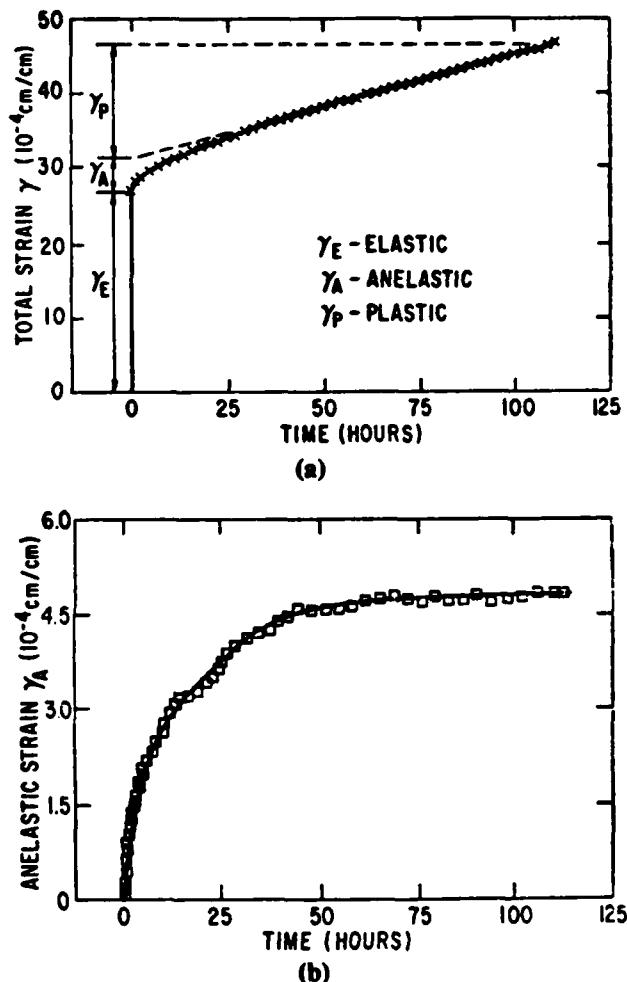


Figure 1. Strain response to equivalent shear stress increase from 105 to 156 MPa. Sample preannealed at testing temperature of 500 K for 325 hours. Symbols \times and \square are actual data. Solid lines are least-square curve fits using the five exponential decay formulation described in the text.

- (a) Total equivalent shear strain ($\gamma = \sqrt{3}$ e-uniaxial tensile strain)
(b) Equivalent anelastic shear strain γ_A , obtained by subtracting the ideal elastic and viscoelastic strains.

Ideal Elasticity

The elastic response of specimens preannealed at 500 K for 325 hours was measured at 293, 424 and 500 K. In all cases, the response was instantaneous, completely reversible, and linear.

The tensile elastic modulus (i.e., Young's modulus) at 293 K was determined to be 84 ± 8 GPa. This is in good agreement with Davis'⁽¹¹⁾ value for Pd₄₀Si₂₀ of 88 GPa.

Using the values of the moduli determined at all three temperatures and assuming a linear temperature dependence, we found that $d(\ln E/dT = (-3.2 \pm 0.6) \times 10^{-4} \text{C}^{-1}$. This is to be compared with the Parry and Pritchett⁽¹⁵⁾ dynamic measurement of $-2.9 \times 10^{-4} \text{C}^{-1}$.

Structural changes in the amorphous state have been shown to affect the elastic stiffness of metallic glasses.⁽²⁾ Increases in modulus approaching 10%, relative to the as-cast condition, have been reported for many systems annealed near the glass transition temperature.⁽³⁾ In our tests, the 325-hour preanneal at 500 K was sufficient to stabilize the structure, and no changes in modulus were observed during the loading and unloading tests.

Anelasticity

Figure 1b shows the anelastic component of the total strain of Figure 1a, obtained by subtracting the ideal elastic and viscoelastic contributions. The time-dependent, transient nature of the flow is evident. Additional anelastic flow is not resolvable after approximately two days.

Complete recoverability is a requirement for true anelastic behavior. In Figure 2, the results of a test of this condition are examined by observing the response to a stress cycle from 35 to 71 to 35 MPa. The entire response is plotted in Figure 2a. The elastic response is seen to be completely and instantaneously recovered on stress reduction ($\gamma_E = -\gamma'_E$). In Figure 2b, only the anelastic contribution is shown. Complete recoverability of this strain response is verified ($\gamma_A = -\gamma'_A$).

The linearity condition for true anelastic flow requires that the total anelastic strain, after the complete decay of the transient, be directly proportional to the magnitude of the stress change. To check this requirement, a creep sample was subjected to sequential stress increases at 500 K, and the anelastic strain was determined in the manner illustrated in Figure 1. In Table 1, the observed anelastic strain is listed for the stress increments, and a linear relation is verified. $\gamma_A/\Delta\tau = (8.0 \pm 0.80) \times 10^{-6} \text{ cm/cm-MPa}$, where τ is the equivalent shear stress $\tau = \sigma/\sqrt{3}$ and σ is the uniaxial tensile stress.

A corollary of the linearity postulate is the superposition (additivity) of the responses.⁽¹⁾ In another load cycle experiment, this principle was checked. The stress was raised in two steps from

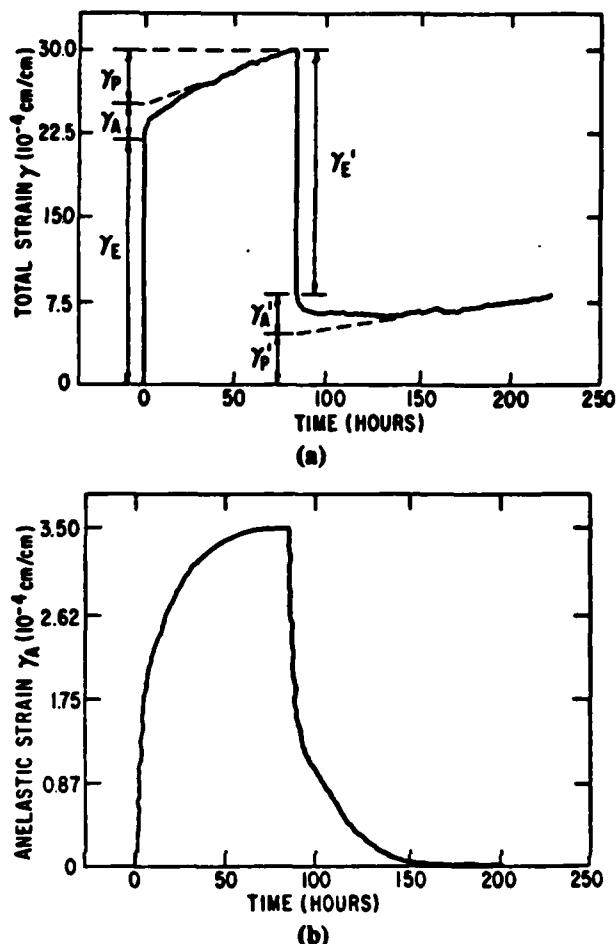


Figure 2. Strain response to stress cycle from 35 to 71 to 35 MPa. Sample preannealed at testing temperature of 500 K for 575 hours.

- (a) Total strain
- (b) Anelastic strain.

Table 1
TOTAL ANELASTIC STRAIN
FOR VARIOUS STRESS INCREASES*

Equivalent Shear Stress Range $\tau = \sigma/\sqrt{3}$ (MPa)	Total Anelastic Equivalent Shear Strain $\gamma_A = \sqrt{3}\epsilon_A$ (10^{-4} cm/cm)	$\gamma_A/\delta\tau$ (10^{-6} MPa^{-1})
54 to 105	4.0 ± 0.3	7.8 ± 0.6
105 to 156	4.7 ± 0.3	9.0 ± 0.6
156 to 207	4.2 ± 0.3	8.1 ± 0.6
207 to 259	4.2 ± 0.3	8.1 ± 0.6
259 to 310	3.6 ± 0.7	7.2 ± 1.2
310 to 363	3.8 ± 0.3	7.2 ± 0.6
363 to 383	1.7 ± 0.3	8.4 ± 1.8

*Specimen preannealed at testing temperature of 500 K for 343 hours.

35 to 71 to 106 MPa. The associated anelastic strains were $(3.5 \pm 0.2) \times 10^{-4}$ cm/cm and $(3.8 \pm 0.3) \times 10^{-4}$ cm/cm. The stress was then reduced in one step to 35 MPa, with a corresponding anelastic recovery of $(7.0 \pm 0.3) \times 10^{-4}$ cm/cm. The superposition of the responses was thus verified.

Although several investigators have examined the effect of structural changes on the anelastic relaxation spectrum of metallic glasses,^(2,4) no systematic investigation of the effect of structural change on the anelastic creep response of these glasses has been reported. We observed no change in the anelastic response of the preannealed specimens during the testing, indicating that the 325-hour preanneal at the testing temperature sufficiently stabilized the structure. It is possible that the anelastic creep response changes during the early stages of annealing, but we have not made observations in this regime.

Viscoelasticity

The viscoelastic flow exhibited by well-annealed samples (i.e., no structural relaxation occurring during the test) is characterized by a constant strain rate. Referring to Figure 1, the viscoelastic component is shown to contribute throughout the test, although constant strain rate steady-state flow is not fully established until after the decay of the anelastic transient.

Viscoelastic flow must, by definition, produce a permanent strain. The test shown in Figure 2a can be used to check this requirement. The total viscous flow γ_p , up to 84 hours (when the stress was reduced), is $(5.0 \pm 0.3) \times 10^{-4}$ cm/cm. The observed permanent set γ'_p is $(5.7 \pm 0.5) \times 10^{-4}$ cm/cm, verifying nonrecoverability of the flow.

The linearity of the viscoelastic stress-strain rate relation has been discussed elsewhere.⁽¹⁶⁾ A transition from linear to nonlinear behavior was shown to be an inherent property of metallic glasses. For Pd₈₂Si₁₈ specimens preannealed and tested under the same condition as those discussed in this study, the limiting condition for linear flow was established at an equivalent shear stress $\tau = \sigma/\sqrt{3} = 300$ MPa, where σ is the uniaxial tensile stress.

The temperature dependence of the isoconfigurational viscoelastic strain rate has been measured. The results are reported in terms of the viscosity $\eta = \tau/\dot{\gamma}_p$ (see Figure 3). To stabilize the structure, the specimens were preannealed under an argon atmosphere with no applied stress. The annealing temperatures and times are indicated. These preanneals enabled us to test the specimens

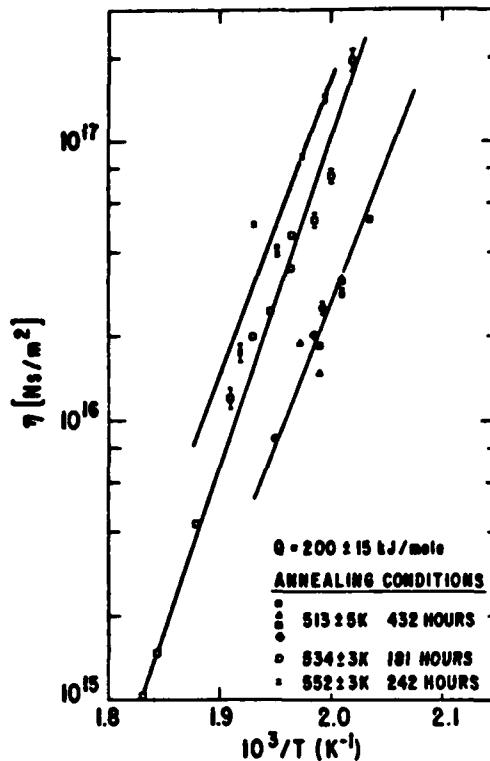


Figure 3. Isoconfigurational viscosities obtained after the indicated annealing treatments.

in the temperature range 490 to 525 K without observing additional structural relaxation. The absence of structural change was verified in each case by returning to the initial testing temperature at the end of the temperature cycle and observing no significant change from the initial value of the viscosity. The activation energies found for the isoconfigurational viscosity of these specimens is 200 ± 15 kJ/mole, in agreement with other measurements for specimens preannealed at lower temperatures.^(9,17) Tests at different stresses also showed no change in the activation energy.

Although the activation energy for isoconfigurational flow is stress and thermal history independent, the magnitude of the viscoelastic strain rate is highly dependent on the degree of structural relaxation, and therefore on thermal history. Previous investigations^(9,17) have shown that the viscosity can be changed by many orders of magnitude, even when annealing at temperatures as low as $T_g - 200$. Figure 4 shows viscosity-time histories for specimens tested under different stresses at 500 K from the cast state (curves A, B, and C). The linear increase of viscosity with time has previously been observed.^(17,18) Note that for the stress range tested ($39 < \tau < 155$ MPa), the viscosity annealing kinetics are stress-independent. Further-

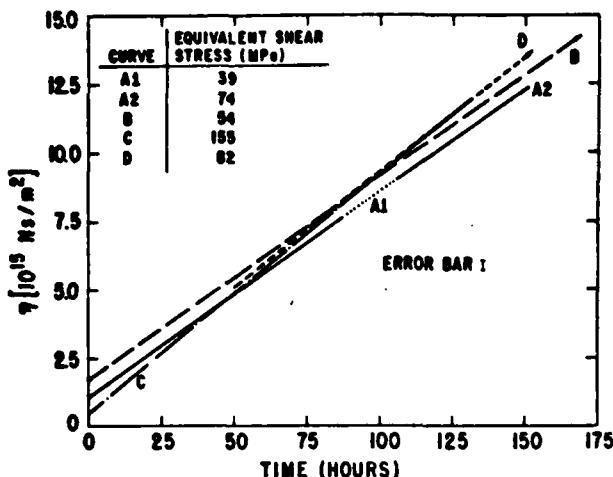


Figure 4. The viscosity as a function of annealing time for specimens tested at the indicated stresses. Samples A, B, and C were creep tested from the as-cast state. The stress on sample A was increased after 80 hours of testing. Sample D was preannealed as described in the text. The origin for sample D was shifted to $t = 50$ hours.

more, there is evidence that subsequent annealing kinetics are not affected by previous thermal history. Curve D is for a specimen that was sequentially preannealed at 424 K for 225 hours, 444 K for 175 hours, 466 K for 152 hours, and 487 K for 240 hours. The specimen was then creep tested at 502 K. During the test, the viscosity was found to increase linearly with time from an initial value of 5×10^{15} Ns/m². In Figure 4, the viscosity-time plot for this test is shown with the origin shifted to coincide with the other curves. The subsequent annealing kinetics are seen to be the same as those for the as-cast specimens, implying that the kinetics of structural relaxation are independent of the thermal history.

For certain properties, such as the Curie temperature,⁽¹⁹⁾ a "cross over effect" is observed during structural relaxation, indicating a thermal history memory. On the other hand, the viscosity does not appear to exhibit this property.

DISCUSSION

Anelastic flow in many crystalline alloys can be characterized as a single process with the general form of an exponential decay: $\gamma_A = \gamma_A^0(1 - e^{-t/\tau})$.⁽¹⁾ We attempted to fit the flow observed in these tests with this relation combined with a viscoelastic term $\gamma(t) = \gamma_A^0(1 - e^{-t/\tau}) + \dot{\gamma}_P t$. A typical least-squares fit, using γ_A^0 , τ , and $\dot{\gamma}_P$ as free parameters, is shown in Figure 5. Only the anelastic strain is shown, the viscoelastic strain $\dot{\gamma}_P t$ having been subtracted from the total strain. As a result of this

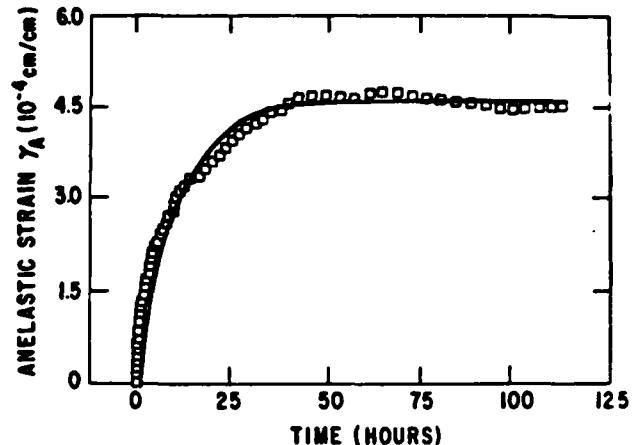


Figure 5. The equivalent anelastic shear strain computed from the data of Figure 1a, using a single exponential decay formulation. The symbols are the actual data points, from which the fitted viscoelastic strain has been subtracted. The solid line is the least-squares fit.

subtraction, the anelastic data peaks at approximately 70 hours and then drops. This is physically unrealistic, and is only an artifact of the fit value for the viscoelastic strain rate $\dot{\gamma}_P$ being higher than the actual value. In addition, the data starts higher than the fitted curve and then crosses it twice before peaking. Similar systematic deviations were observed in every case examined in this manner. This incompatibility of the data with a single process exponential decay is an indication of the existence of an anelastic site spectrum. The existence of such a spectrum in the more common glassy systems is well established.⁽²⁰⁾ Recent work on metallic glasses has shown that anelastic flow in these materials is also governed by a spectrum.^(8,15) The anelastic flow can then be expressed as $\gamma_A = \sum_i \gamma_{Ai}^0(1 - e^{-t/\tau_i})$, where the sum is replaced by an integral for the case of a continuous spectrum. Fitting experimental data by an exact method to such a relation is difficult, and one must resort to an approximation technique.⁽²¹⁾ We chose the finite spectrum approach, employing a least-squares fit to the prefactor of five exponential decays, with predetermined time constants spaced by a factor of three ($\tau_i = 0.2, 0.6, 1.8, 5.0$, and 17 hours). The smallest time constant is limited by the data acquisition rate, one point every 0.1 hour. The decay with the largest time constant, 17 hours, reaches 0.95 of its final value after 50 hours. From the recovery tests, it is known that additional anelastic flow is undetectable after this time. Therefore, this spectrum spans the necessary time constants. Performing a least-squares fit to the data

with this relation combined with a viscoelastic term $\gamma(t) = \sum_i \gamma_i (1 - e^{-t/\tau_i}) + \dot{\gamma}_p t$ and using γ_i and $\dot{\gamma}_p$ as free parameters, we produced an excellent fit to the data (see the solid lines in Figure 1). Note that the fit value for the viscoelastic strain rate $\dot{\gamma}_p$ is reasonable since it does not produce an artificial peak in the anelastic strain. This method provides an analytic expression from which the total anelastic flow can be computed. However, fitting the data with a finite sum of exponentials whose time constants are arbitrarily predetermined, yields a nonunique solution, and therefore no physical significance could be attributed to the weights of the resulting spectrum.^(1,21)

CONCLUSIONS

The homogeneous strain response of amorphous Pd₈₂Si₁₈ to an applied stress consists of three components: elastic, anelastic, and viscoelastic strains.

The elastic response is fully reversible, instantaneous, and linear. The tensile elastic modulus at 293 K was determined to be 84 ± 8 GPa. The temperature dependence is $d(\ln E)/dT = (-3.2 \pm 0.6) \times 10^{-4} \text{C}^{-1}$.

The anelastic response is a transient that follows stress increases and is fully recovered upon stress reduction. For a Pd₈₂Si₁₈ specimen, preannealed and tested at 500 K, the anelastic transient contributes measurably of the flow for approximately 50 hours. The transient cannot be described by a single process exponential decay. However, a sum of exponential decays spanning a spectrum of time constants from 0.2 to 17 hours provides an excellent fit to the data.

The total anelastic strain, after the complete decay of the anelastic transient, varies linearly with the magnitude of the stress change over the entire stress range tested: $\gamma_a/\Delta\tau = 8.0 \pm 0.8 \times 10^{-6} \text{ MPa}^{-1}$. The superposition principle was shown to apply to anelastic flow.

The isostructural viscoelastic flow is nonrecoverable and characterized by a constant strain rate. The activation energy for isoconfigurational flow was found to be 200 ± 15 kJ/mole, independent of the annealing temperature used to stabilize the structure. When structural relaxation occurs, the viscoelastic strain rate changes by many orders of magnitude. The strain rate has been observed to vary inversely with time. The rate of decrease was found to be both stress and thermal history independent.

The viscoelastic strain rate-stress relation is linear only in the low stress regime, a transition to nonlinear behavior occurring at approximately

300 MPa. The anelastic strain, on the other hand, remains linear over the entire stress range tested (0 to 383 MPa).

ACKNOWLEDGMENT

We want to thank Professor D. Turnbull for many helpful suggestions and Dr. B.S. Berry for several useful discussions. A.I. Taub gratefully acknowledges the financial support provided by an IBM Predoctoral Fellowship. The experimental work was performed at Harvard University and supported by the Office of Naval Research under Contract N00014-77-C-0002.

REFERENCES

1. A.S. Nowick and B.S. Barry, *Anelastic Relaxation in Crystalline Solids*, Academic Press, New York, 1972.
2. H.S. Chen, *Rep. Prog. Phys.* 43 (1980) 353.
3. H.S. Chen, *J. Appl. Phys.* 49 (1978) 3289.
4. B.S. Berry, *Metallic Glasses*, American Society of Metals, Cleveland, Ohio, 1978, p. 161.
5. R. Maddin and T. Masumoto, *Mat. Sci. Eng.* 9 (1972) 153.
6. J. Logan and M.F. Ashby, *Acta Met.* 22 (1974) 1047.
7. T. Murata, H. Kimura, and T. Masumoto, *Scripta Met.* 10 (1976) 705.
8. A.S. Argon and H.Y. Kuo, *J. Non-Cryst. Solids* 37 (1980) 241.
9. A.I. Taub and F. Spaepen, *Scripta Met.* 13 (1979) 195.
10. H.S. Chen and M. Goldstein, *J. Appl. Phys.* 43 (1971) 1642.
11. L.A. Davis, *Metallic Glasses*, American Society of Metals, Cleveland, Ohio, 1978, p. 190.
12. F. Spaepen, *Acta Met.* 25 (1977) 407.
13. A.S. Argon, *Acta Met.* 27 (1979) 47.
14. A.I. Taub and F. Spaepen, *Scripta Met.* 13 (1979) 883.
15. B.S. Berry and W.C. Pritchett, *J. Appl. Phys.* 44 (1973) 3122.
16. A.I. Taub, *Acta Met.* 28 (1980) 633.
17. A.I. Taub and F. Spaepen, *Acta Met.* 28 (1980) 1781.
18. P.M. Anderson III and A.E. Lord, *Mat. Sci. and Eng.* 44 (1980) 279.
19. A.L. Greer, *Thermochimica Acta*, to be published.
20. J.D. Ferry, *Viscoelastic Properties of Polymers*, John Wiley, New York, 1970.
21. C. Lanczos, *Applied Analysis*, Prentice-Hall, Englewood Cliffs, New Jersey, 1956, Chap. 4.

Defense Documentation Center
Cambridge Station
Alexandria, Virginia 22314 (12)

Office of Naval Research
Department of the Navy
Attn: Code 471
Code 105
Code 470 (3)

Director
Office of Naval Research
Branch Office
495 Summer Street
Boston, Massachusetts 02210 (6)

Director
Office of Naval Research
Branch Office
936 South Clark Street
Chicago, Illinois 60607

Office of Naval Research
San Francisco Area Office
700 Market Street, Room 447
San Francisco, California 94102

Naval Research Laboratory
Washington, D.C. 20390
Attn: Code 4000
Code 6100
Code 6100
Code 6100
Code 2627 (6)

Attn: Mr. F. S. Williams
Naval Air Development Center
Code K2
Warminster, Pennsylvania 18974

Naval Air Propulsion Test Center
Trenton, New Jersey 08628
Attn: Library

Naval Weapons Laboratory
Dahlgren, Virginia 22448
Attn: Research Division

Naval Constructors Battalion
Civil Engineering Laboratory
Port Hueneme, California 93043
Attn: Materials Division

Naval Electronics Laboratory Center
San Diego, California 92132
Attn: Electronic Materials Sciences Div.

Naval Missile Center
Materials Consultant
Point Mugu, California 93041
Commanding Officer

Naval Ordnance Laboratory
White Oak
Silver Spring, Maryland 20910
Attn: Library

Naval Ship R. & D. Center
Materials Department
Annapolis, Maryland 21402

Naval Undersea Center
San Diego, California 92132
Attn: Library

Naval Underwater System Center
Newport, Rhode Island 02840
Attn: Library

Naval Warfare Center
China Lake, California 93555
Attn: Library

Naval Postgraduate School
Monterey, California 93940
Attn: Materials Sciences Dept.

Naval Air Systems Command
Washington, D.C. 20380
Attn: Code 52031
Code 52032
Code 320

Naval Sea System Command
Washington, D.C. 20384
Attn: Code 035

Naval Facilities
Engineering Command
Alexandria, Virginia 22331
Attn: Code 03

Scientific Advisor
Commandant of the Marine Corps
Washington, D.C. 20380
Attn: Code AX

Naval Ship Engineering Center
Department of the Navy
Washington, D.C. 20380
Attn: Director, Materials Sciences

Army Research Office
Box 124, Duke Station
Durham, North Carolina 27709
Attn: Materials and Ceramics Div.

Army Materials and Mechanics
Research Center
Watertown, Massachusetts 02472
Attn: Res. Programs Util-JAMXMA-P

Commanding General
Department of the Army
Frankford Arsenal
Philadelphia, Pennsylvania 19137
Attn: ORDBA-1320

Office of Scientific Research
Department of the Air Force
Washington, D.C. 20331
Attn: Solid State Div. (SRPG)

Aerospace Research Lab
Wright-Patterson AFB
Building 450
Dayton, Ohio 45433

Air Force Materials Lab (LA)
Wright-Patterson AFB
Dayton, Ohio 45433

NASA Headquarters
Washington, D.C. 20586
Attn: Code 2304

NASA
Lewis Research Center
21000 Brookpark Road
Cleveland, Ohio 44135
Attn: Library

National Bureau of Standards
Washington, D.C. 20234
Attn: Metallurgy Division
Inorganic Materials Division

Atomic Energy Commission
Washington, D.C. 20235
Attn: Metals and Materials Branch

Defense Metals and Ceramics
Institute on Ceramics
Battelle Memorial Institute
505 King Avenue
Columbus, Ohio 43201

Director
Ordnance Research Laboratory
P.O. Box 32
State College, Pennsylvania 16801

Director Applied Physics Lab
University of Washington
1013 Northeast Fortis Street
Seattle, Washington 98103

Metals and Ceramics Division
Oak Ridge National Laboratory
P.O. Box X
Oak Ridge, Tennessee 37830

Los Alamos Scientific Lab.
P.O. Box 1663
Los Alamos, New Mexico 87544
Attn: Report Libraries

Argonne National Laboratory
Metallurgy Division
P.O. Box 229
Lemont, Illinois 60439

Brookhaven National Laboratory
Technical Information Division
Upton, Long Island
New York 11973
Attn: Research Library

Library
Building 50, Room 134
Lawrence Radiation Laboratory
Berkeley, California

Professor G. S. Ansell
Rensselaer Polytechnic Institute
Dept. of Metallurgical Engineering
Troy, New York 12181

Professor H. K. Birnbaum
University of Illinois
Department of Metallurgy
Urbana, Illinois 61801

Dr. E. M. Breiden
United Aircraft Corporation
United Aircraft Research Lab
East Hartford, Connecticut 06108

Professor H. D. Brody
University of Pittsburgh
School of Engineering
Pittsburgh, Pennsylvania 15213

Professor J. B. Cohen
Northwestern University
Dept. of Material Sciences
Evanston, Illinois 60201

Professor M. Cohen
Massachusetts Institute of Technology
Department of Metallurgy
Cambridge, Massachusetts 02139

Professor B. C. Grossen
Northwestern University
Department of Chemistry
Boston, Massachusetts 02115

Dr. G. T. Hahn
Battelle Memorial Institute
Department of Metallurgy
515 King Avenue
Columbus, Ohio 43211

Professor F. W. Herzel
Carnegie-Mellon University
Schenley Hall
Pittsburgh, Pennsylvania 15213

Dr. David C. Hoeden
Battelle Memorial Institute
Columbus Laboratories
505 King Avenue
Columbus, Ohio 43201

Professor C. E. Jacobs
Ohio State University
Dept. of Welding Engineering
190 West 19th Avenue
Columbus, Ohio 43210

Professor C. Judd
Rensselaer Polytechnic Institute
Dept. of Materials Engineering
Troy, New York 12181

Dr. C. S. Koertenbach
TRW, Inc.
3355 Euclid Avenue
Cleveland, Ohio 44117

Professor D. A. Koss
Michigan Technological University
College of Engineering
Houghton, Michigan 49931

Professor A. Lawley
Drexel University
Dept. of Metallurgical Engineering
Philadelphia, Pennsylvania 19104

Dr. N. Margolin
Polytechnic Institute of New York
333 Jay Street
Brooklyn, New York 11201

Professor K. Massabumi
Massachusetts Institute of Technology
Department of Ocean Engineering
Cambridge, Massachusetts 02139

Dr. G. H. Meyer
University of Pittsburgh
Dept. of Metallurgical and Materials
Engineering
Pittsburgh, Pennsylvania 15213

Professor J. W. Morris, Jr.
University of California
College of Engineering
Berkeley, California 94720

Professor K. Ooe
University of California
Materials Department
Los Angeles, California 90024

Professor W. F. Savage
Rensselaer Polytechnic Institute
School of Engineering
Troy, New York 12181

Dr. C. Shaw
Rockwell International Corp.
P.O. Box 1165
1649 Camino Del Rio
Thousand Oaks, California 91360

Professor O. D. Sherry
Stanford University
Materials Sciences Dept.
Stanford, California 94300

Professor J. Shynk
Stanford University
Materials Sciences Department
Stanford, California 94300

Dr. W. A. Spring
U.S. Steel Corporation
Research Laboratory
Monroeville, Pennsylvania 15146

Dr. E. A. Smakle, Jr.
Georgia Institute of Technology
School of Chemical Engineering
Atlanta, Georgia 30332

Professor H. S. Stollhoff
Rensselaer Polytechnic Institute
School of Engineering
Troy, New York 12181

Dr. E. R. Thompson
United Aircraft Research Lab.
400 Main Street
East Hartford, Connecticut 06108

Professor David Turnbull
Harvard University
Division of Engineering and Applied
Physics
Cambridge, Massachusetts 02139

Dr. F. W. Wang
Naval Ordnance Laboratory
Patuxent Laboratory
Antietam Lab
Silver Spring, Maryland 20910

Dr. J. C. Williams
Rockwell International
Defense Center
P.O. Box 1165
Thousand Oaks, California 91360

Professor H. G. F. Wieder
University of Virginia
Department of Materials Science
Charlottesville, Virginia 22903

Dr. M. A. Wright
University of Tennessee
Space Institute
Dept. of Metallurgical Engineering
Tennessee, Tennessee 37382